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Research in the Iron and Steel Industry.

WE have recently discussed the steps that are being taken to ensure adequate technical development within the industry as time proceeds, but the recent opening by Lord Rutherford of the United Steel Company's new research laboratories fully justifies a return to the subject. This latest addition to the already existing series of research laboratories acting for particular groups of works must be welcomed as an example, not only of the best modern outlook on the part of those in control of the company's policy, but also as an indication of their optimism as regards the future of the industry. The equipment and staff of the laboratories leaves nothing to be desired, whilst the technical field to be dealt with covers mainly the whole of the diverse activities of the heavy steel trade, as expressed in the chain of processes and products coincident with the associated works of the parent company.

The development and strengthening of research organisations within the industry as in this case is the necessary corollary to the centralised efforts of the British Federation of Iron and Steel Manufacturers in its technical organisation and committee groupings, which co-ordinated effort is being so materially assisted by the Department of Scientific and Industrial Research. Effort in this direction is not likely to be overdone. The need is great, and the money is apparently forthcoming. The question now to be answered is whether we have available a sufficient number of the right type of adequately trained men. The times of adversity through which our own industries, in common with those of other countries, have passed, for a time placed them in a disadvantageous position as regards the absorption of men from our universities and colleges. With the resuscitation of our national industries, there must necessarily develop an accentuated demand for the highly trained physicist and chemist, and it is to be hoped that many more of our brilliant young men will seek their future in the industrial field by this avenue.

It has been said by Professor E. W. Hobson, F.R.S., that when problems for research have been exhausted, "completely rationalised physics and chemistry would contain within themselves, in the form of postulates, every element which could be supplied by physical observation, and would no longer be dependent for their future progress upon the work of the experimenter. Laboratories would then be useful only for illustrative, didactic, and suggestive purposes, just as drawings and models are still needed in geometry, but they would no longer hold their present indispensable position in relation to research." The conditions of the Professor's delightful postulation are still many hundreds of years away from fulfilment. There still awaits the experimenter an infinity of possibilities; possibilities as regards process as well as product.

Too many of our processes are approached through the traditional grooves; it needed the advent of the seaplane seriously to modify the standard distance between the wheels of the Roman chariot. Only during the last few months have we learned to our great surprise that a magnet of extremely outstanding properties may be produced from an iron-nickel-aluminium alloy, a great recent achievement of Japanese metallurgy. There are, indeed, still great possibilities. Professor Hobson's conditions are *not* as yet fulfilled.

Rimmed Steel Ingots.

IN the manufacture of rimmed steel, the molten metal is not deoxidised, or is only partially deoxidised, before it is poured into the moulds. Reactions occur during solidification between free oxides, of which the chief is probably iron oxide, and carbon in the metal as a result of which carbon and oxygen are partially eliminated in the form of gas. This gas rises to the top of the solidifying ingot, and, together with other gases which are liberated during the solidification process, causes the metal at the top of the ingot to boil, while freezing of the metal proceeds inwards from the mould walls with the formation of a gradually increasing rim. As a rule, the rimming action is not allowed to proceed to completion, but the molten metal at the top centre of the ingot is chilled, and further elimination of the gas checked, by covering the top with a plate or cast iron block. Various factors affect the quality of rimmed steel, such as composition, degree of segregation, and internal structure, and these are considered, as well as the production of the steel more especially for the manufacture of deep drawing sheets and strip, by J. H. Nead and T. S. Washburn, in the March issue of *Metals and Alloys*.

The ladle composition which has been found to be most satisfactory for deep drawing steel is: Carbon, 0·07 to 0·10 per cent.; manganese, 0·35 to 0·45 per cent., with maximum values of 0·015 per cent. and 0·030 per cent. for phosphorus and sulphur respectively. In the ingot the carbon is about 0·03 per cent. lower as a result of the rimming action, phosphorus is normally less than 0·010 per cent., and sulphur should be less than 0·025 per cent. A clean scrap charge—that is, a charge consisting of heavy and light melting mill scrap—has been found to produce better quality steel, in respect of both surface and drawing properties, than charges of miscellaneous scrap. Again, it has been found that the use of limestone yields steel freer from laminations than the use of burnt lime.

With regard to the effect of the composition of

the iron on the quality of the finished steel, a satisfactory range has been found to be 0·75 to 1·35 per cent. silicon, 1·50 to 2·50 per cent. manganese, 0·150 to 0·250 per cent. phosphorus, and 0·040 per cent. sulphur, although it is preferable for the sulphur to be less than 0·030 per cent. The amount of lime which is used in the charge depends on the amount of silicon in the iron and scrap and on the iron oxide content of the slag. The iron content of the slag increases with the basicity or lime-silica ratio, and for a given basicity the iron oxide of the slag increases with decreasing carbon content. Satisfactory rimming action was obtained in heats of 0·07 to 0·10 per cent. carbon content, with an iron oxide content in the slag between 17 and 21 per cent. This was maintained by a lime-silica ratio between 3·0 and 3·3 per cent., which necessitated a lime charge of 9·5 per cent. When the required carbon content has been obtained, spiegel is added to the bath, which is tapped thirty minutes afterwards. The effect of the spiegel is to equalise the temperature of the bath and flux out some of the impurities. Aluminium in stick form is also added to the ladle in amounts depending on the iron oxide content of the slag before the spiegel addition. The aluminium addition tends to compensate for variations in the iron oxide content of the slag, and helps to control the rimming action. The desired condition in the mould is for the metal level to sink slightly after topping off, and before rimming commences, for this has been found to correspond with the most satisfactory ingot structure for deep-drawing steel.

Split ingot sections were obtained for examination by drilling holes 1¾in. in diameter, 6in. apart along the centre line of one side, half-way through the ingot and filling them with suitable charges, which were then electrically detonated. In rimmed steel ingots there is no well-defined pipe, such as there is in normal, killed ingots, but there is, instead, a corresponding porous zone in which carbon, phosphorus, and sulphur are segregated. Other characteristic structural features of rimmed steel ingots are a primary blow-hole zone, extending along the base and about halfway up the ingot, and a secondary blow-hole zone some inches from the surface and extending almost along the length of the ingot. In the type of ingot most satisfactory for the manufacture of deep drawing steel, the primary blow-holes are almost suppressed, the secondary blow-holes, although well defined, are deep-seated, and there is relatively little carbon concentration in the upper portion. Ingots of this type are only obtained with satisfactory open-hearth practice, and when ingots are of the correct size, for it has been found that there is a limit to the height for an ingot of given cross section beyond which it is difficult to obtain the required ingot structure.

An abrupt change in composition occurs at the secondary blow-hole zone which results from the change in ingot solidification consequent upon the capping of the ingot and the cessation of the rimming action. The steel between the zone of secondary blow-holes and the surface is lower in carbon content than the central zone because of selective freezing. As soon as the rimming action ceases, normal segregation of carbon occurs towards the centre and upper portion of the ingot. Tests on two ingots showed, for a steel with a carbon content in the ladle of 0·097 per cent., an average carbon content of 0·040 per cent. in the zone between the secondary blow-holes and the surface, and 0·064 per cent. at the core. While phosphorus and sulphur were distributed similarly to the carbon, there was little difference in the manganese content of the two zones, although it was segregated to some extent at the top of the inner zone. Ingots which have been found satisfactory for the production of deep drawing sheets have the following characteristics :—Small primary blow-holes, the outer ends of which are ¼in. or more from the surface at the bottom of the ingot, and which become progressively deeper until they disappear about halfway up the ingot ; a deep-seated secondary blow-hole zone composed of small blow-holes ; and a minimum segregation of carbon, phosphorus, and sulphur in the upper half of the inner zone.

On the Production of Aluminium-Copper "Hardeners."

By EDMUND R. THEWS.

IT is frequently necessary in the industrial production of alloys to add certain alloying components in the form of "hardeners" or primary alloys, serving either to introduce metals of high melting points in an easily soluble state or to prevent excessive loss of alloying metals by oxidation or volatilisation. Alloys of the first type are the red and white alloys with copper, manganese, nickel, &c., whilst the easily oxidised or vapourised elements of the second type are most commonly represented by phosphorus, zinc, tin, magnesium, silicon, &c.

Among the various metals requiring the addition of high melting alloying components in the form of primary alloys, aluminium is the most prominent. On account of its relatively low melting temperature, its high affinity for atmospheric oxygen, and its absorptive capacities for atmospheric nitrogen, and other gases at comparatively low temperatures, direct addition of high melting and slowly alloying metals would lead to serious contamination of the aluminium alloys with oxides and gases. In a number of instances there is also the danger that on account of unfavourable alloying properties absolutely satisfactory alloys cannot be obtained under fairly normal foundry conditions if such high melting alloys are directly added to the aluminium, since even if the alloying additions are softened, it is rather difficult under practical working conditions to produce perfectly homogeneous castings. If, however, these high melting metals are first combined with aluminium under suitable conditions, melting points are reduced and general alloying properties improved to such an extent that the production of the final alloy can be accomplished with comparative ease.

Most of the hardeners used in the production of aluminium alloys are of the binary type, consisting besides aluminium of the high melting or easily oxidised or volatilised metal in question. If a number of such metals—such as copper and iron—are to be added to the final metal at the same time, they may conveniently be introduced by one ternary or even quaternary hardener, although special conditions may occasionally render it advisable to employ a number of separate binary hardeners.

An instance of the latter type is the addition of light bronze or brass materials to the aluminium wherever the final alloys are to contain tin or zinc besides copper. New and clean sheets or strips are usually employed for this purpose. Attention should in this connection be called to the fact that while the addition of clean brass sheets and strips to aluminium

may under well-controlled melting conditions yield high grade aluminium-copper-zinc alloys, less satisfactory results are usually obtained with corresponding bronze materials for the production of aluminium-copper-tin mixtures. It is obvious that this difference is not bound up with melting temperatures, since many bronze alloys melt at lower temperatures than the customary brasses, and it appears that this difficulty is rather due to unfavourable solubility phenomena. This supposition is supported particularly by the fact that aluminium-copper-tin alloys produced by the addition of light bronze material to the super-heated aluminium rarely or never contain hard or unmelted particles, the heterogeneity of the final alloy extending merely to non-uniform mixing of the fused bronze particles with the aluminium. It is possible, of course, to obtain absolutely homogeneous alloys by melting and alloying aluminium and bronze in a vacuum, permitting the use of higher temperatures and more intimate mixing, but since apparatus of this type is rarely available, aluminium-copper-tin alloys should be produced by adding the copper in the form of an aluminium-copper hardener and the requisite quantities of tin in its unalloyed state immediately before pouring.

The most common binary hardeners for aluminium are :—Aluminium-copper, aluminium-silicon, aluminium-manganese, aluminium-iron, and aluminium-nickel, the most frequently used ternary alloys being aluminium-copper-iron and aluminium-copper-manganese.

Hardeners are usually melted in graphite crucibles, although much better results are obtainable with suitable electric furnaces of the indirect heating type. Deep-hearth reverberatory furnaces and cast iron melting pots are also employed, but these form exceptions to the rule. Reverberatories are advantageous only where daily capacities of 15 to 25 tons are required, their chief advantage over smaller melting units being the much higher degree of uniformity of the products obtained. It is obvious that if, say, 10 tons of aluminium-copper with 40 per cent. copper are melted in one single reverberatory charge, the entire quantity can with some care be made of practically uniform composition, whilst if this amount of alloy is melted in forty crucible charges of 250 lb. each, the copper contents of the individual batches are bound to fluctuate considerably. However, since normal production quantities rarely exceed a few tons a day, crucible furnaces are the common melting apparatus employed for this purpose.

In certain instances—that is to say, where the alloying metals are poured together—a number of furnaces are used simultaneously. In producing aluminium-copper hardeners, crucible furnaces are usually employed for melting copper as well as aluminium, although occasionally the aluminium is melted in cast iron melting pots, crucible furnaces being employed exclusively for the melting of copper.

Commercial aluminium-copper hardeners are produced in three qualities, alloys with 33, 40 and 50 per cent. copper respectively. Of these, the latter composition is most commonly used, since its outstanding advantage, the carrying of maximum percentages of copper, is not accompanied by any serious disadvantage as compared to the other less concentrated alloys. The 33 per cent. alloy is occasionally claimed to be more advantageous in that it melts at the lowest temperature of the three compositions mentioned, *viz.*, at 540 deg. Cent. (1000 deg. Fah.) without exhibiting pronounced segregation tendencies. However, the difference between the melting points of the 33 per cent. alloy and the 50 per cent. mixture amounts to about 35 deg. Fah. only, the absolute melting point of the latter alloy, 575 deg. to 580 deg. Cent. (1065 deg. to 1075 deg. Fah.), still being sufficiently far below that of aluminium to ensure rapid and perfect melting of this hardener at the ordinary working temperatures. The still lower melting point of the 33 per cent. alloy is of no practical importance, therefore. The same applies to the tendency of the 50 per cent. aluminium-copper alloy towards segregation, which is of no greater practical consequence than that of the 33 per cent. mixture. In any case, if melting and casting of this alloy are carried out with due care, the homogeneity of the product is such that any particular piece of walnut size will have the same composition as the alloy in general.

Aluminium-copper hardeners may be produced by the following methods :—

(1) Adding solid aluminium to molten copper ;
(2) Adding solid copper to molten aluminium ;
(3) Adding molten copper to molten aluminium ;
(4) Adding molten aluminium to molten copper.

Of these four possibilities, only the second and third are in common usage. It appears that for general purposes the third process—adding molten copper to molten aluminium—is the most satisfactory, since apart from being the shortest and simplest method of combining the two metals, it gives rise to a smaller loss of aluminium by oxidation than adding solid copper to the molten aluminium (process No. 2). However, the latter process is most generally employed in America, and in connection with special square or cylindrical furnaces of comparatively deep hearths. It appears that this preference is due to the advantages offered by the combination of the method with the special furnaces developed for this purpose rather than to any inherent advantages of the process itself.

The chief advantage of pouring molten copper into the bath of aluminium is the fact that the latter need not be maintained at the elevated temperature required for alloying for an extended period of time, and it appears that the chief characteristic of the American method of dissolving solid copper in superheated aluminium is the maintenance of the aluminium bath under a reducing, or neutralising, atmosphere which protects the metal against atmospheric oxidation and absorption of atmospheric gases.

The production of an aluminium-copper hardener by pouring molten copper into the liquid aluminium exhibits the one disadvantage, however, of inducing a very strong exothermic reaction between aluminium and the commercial grades of copper. This phenomenon has not been definitely explained as yet. The fact exists that on pouring molten copper into molten aluminium the temperature of the bath is raised by about 500 deg. to 600 deg. Fah., which is commonly explained by the exothermic combination of the aluminium with the oxide contents of the copper as well as by the heat of formation of the aluminium-copper alloy.

It appears, however, that the latter influence is of rather small importance in comparison with the exothermic effects of the aluminium-oxygen reaction. For instance, if previous to pouring the copper into the aluminium the former is deoxidised with phosphor copper, the temperature rise of the mixture corresponds rather closely to the surplus of heat calories introduced by the higher melting copper.

However, while the metallurgical disadvantages of this process can be avoided by the deoxidation

of the copper prior to alloying, this precautionary measure is rarely employed under practical working conditions, since this exothermic generation of heat is conveniently employed for melting additional quantities of metal. This is usually carried out by liquifying only about 75 per cent. of the total quantity of aluminium required, adding the rest directly before pouring in the copper. If this final addition of solid aluminium is made in the form of small blocks, the heat generated by the interaction between aluminium and copper will liquify the solid aluminium present in the mixture, the total heat of reduction being lowered accordingly. This method does not take into consideration the very high latent heat of fusion of aluminium, however, and it is advisable, therefore, to add the copper only after the solid aluminium forming the subsequent addition has had sufficient time to absorb practically all the heat available in the aluminium bath. If this latter method is carried out correctly, the molten copper is then poured into a practically semi-liquid, or pastose, bath of aluminium, the heat of reaction generated being immediately absorbed by the liquidation of the pasty aluminium.

It is necessary to point out in this connection that if the copper is thus added to the aluminium in its original state—that is to say, without previous deoxidation of the cuprous oxide contents, the alloy is bound to be contaminated with aluminium oxide, the formation of which is the principal reason for the exothermic reaction between aluminium and copper, some of the oxide formed will rise to the surface of the bath, while another fraction of the total quantity of aluminium oxide produced can be removed by washing the melt with a suitable flux, or by treating it with phosphor copper immediately before pouring. Nevertheless, a certain amount of this aluminium oxide remains with the finished alloy, and while it may not be sufficient to influence the quality of common aluminium alloys to which the hardener is to be added, it often becomes a serious factor where quality alloys are to be produced.

It is absolutely necessary in the latter case, therefore, to deoxidise the copper before alloying, a small excess of phosphor copper being without influence on the commercial usefulness of the hardener. About 0·15 to 0·25 per cent. phosphor copper (with 15 per cent. phosphorus) is added to the copper about five minutes before pouring. The melt is thoroughly stirred and then left to stand quietly to permit the oxide of phosphorus to separate from the copper. The melt is then skimmed and poured into the aluminium.

While the generation of heat by deoxidation of the copper by the aluminium is thus lost to the process, the considerable increase of the aluminium temperature induced by the addition of the hot copper can still be usefully employed in melting additional quantities of aluminium by the process or processes indicated above, although the amount of solid aluminium added must be decreased accordingly. Under ordinary circumstances it is possible to add about 10 or 15 per cent. of the total quantity of aluminium in the form of small aluminium blocks about five minutes before the addition of the copper. It is sufficiently softened by this time to absorb the excess heat of the bath within a minimum of time.

The alternative process of producing aluminium-copper hardener by *adding solid copper to molten aluminium* is carried out by stirring light and thin-walled copper materials, such as clean sheet cuttings, strip or wire cuttings, &c., into the aluminium bath heated to a temperature of about 800 deg. Cent. (about 1450 deg. Fah.). Excessive oxidation of the aluminium by atmospheric oxygen is prevented by covering the bath with a protective layer of flux. A very effective fluxing mixture of this type consists of 85 per cent. of table or sea salt, and 15 per cent. of cryolite. Another satisfactory mixture consists of 1 part of borax and 2 parts of cryolite, while a third mixture recommended for this purpose is composed of 10 parts of calcinated soda (soda ash) and 2 parts of calcinated potash. However, the last mixture should not be employed in connection with graphite crucibles, since it exerts a very corrosive influence on the silicious constituents of the clayey components of the crucible material. The same applies to reverberatory furnaces, which are rapidly corroded by fluxing mixtures of as caustic a nature as the above. The most satisfactory mixture of the three in the author's experience is the sea salt-cryolite, which is as effective as it is simple and stable, although in order to obtain maximum results it is necessary to mix the two constituents as intimately as possible, or, if possible, to grind them together.

If the temperature of the aluminium bath is maintained at about 1450 deg., or, perhaps, 1500 deg. Fah., during the alloying of the thin-walled copper material, and if the mixture is stirred continuously, alloying can be effected within a comparatively short space of time, the resulting alloy—cast in thin slabs—being uniform and homogeneous.

It is obvious from the above that this process is not as satisfactory and efficient as the one first described, its only clear advantage being that the process can be carried out in one melting appliance. It is slower and more tedious than the former method, while a particularly serious disadvantage is the introduction of all the oxide present on and in the copper material added. It is well known that practically all commercial copper material contains a small percentage of cuprous oxide left in the metal during fire refining to prevent excessive absorption of atmospheric gases by the fused copper. Besides, all copper material—even if of clean and bright appearance—has a thin film of oxide spread over its surface, and all this internal and external oxide is introduced into the bath, where it is reduced by the aluminium. It has been pointed out above that by treating the alloy with suitable fluxes (the cryolite mixtures given above being the most efficient), and by stirring in a small percentage of phosphor copper prior to pouring some of the aluminium oxide thus formed, can be removed, but the fact remains that high-grade hardeners of this type should not contain aluminium oxide, or at least not more than traces, in order to be adapted to the production of the highest grade aluminium copper alloys.

Adding solid aluminium to molten copper, therefore, appears to be more satisfactory than the reverse process, since it permits of the deoxidation and purification of the copper before adding the aluminium. The chief difficulty of this process is that if the copper is heated to a temperature just above its melting point in order to prevent overheating of the aluminium, the high latent heat of fusion of the latter will extract so much heat from the mixture as to cause temporary freezing. If, on the other hand, the temperature of the copper is raised sufficiently to preclude this possibility, the resulting aluminium-copper may contain excessive quantities of aluminium oxide. In any case, this process represents a theoretical possibility only, and is rarely, or never used under practical conditions.

The above applies in a still larger degree to the fourth process of production of aluminium-copper hardeners, namely, *the addition of molten aluminium to molten copper*. It is possible, perhaps, to select an alternative of the two last methods calculated to avoid the ill-effects of the latent heat of fusion of the aluminium by adding this metal in a semi-fluid, or pastose, state to the copper heated to a temperature only about 100 deg. above its melting point, but there can be no doubt that either of the two last-named processes possess more disadvantages than advantages, if compared with the pouring of molten copper into molten aluminium.

Age-Hardening Alloys.

IN a paper dealing with the present status of age-hardening in the June issue of the "Transactions" of the American Society for Metals, by R. H. Harrington, three general types of age-hardening are recognised, namely, simple precipitation hardening, simple lattice-strain age-hardening, and complications due to allotropy. The first of these is exemplified by the iron-tungsten alloys containing between 8 and 32 per cent. of tungsten. Quenching alloys of this system in this range of composition from above the solubility line results in the retention of the compound Fe_3W_2 in the alpha solid solution and on subsequent reheating to suitable temperatures an increase in hardness occurs from the precipitation of particles of the compound in fine dispersion. The 15 per cent. tungsten alloy, for example, when oil quenched from 1200 deg. Cent. has a Brinell hardness of 147, which is increased to 230 by subsequent reheating to 800 deg. Cent. Much work has been done on the duralumin type of alloys, but the actual mechanism of the ageing process is not yet completely known. Both the compounds Mg_2Si and $CuAl_2$ are involved. It has been inferred that there is at first an atomic precipitation which strains the lattice and causes an increase in hardness and strength. In some of the alloys the precipitation of a compound or compounds occurs on heating, and masks the lattice-strain hardening. The precise causes which bring about hardening in alloys of this type depend both on the compositions of the alloys and the conditions of ageing. The iron-carbon system affords a good and certainly well-known example of simple precipitation hardening reactions complicated by allotropy. Austenite retained in quenched steels can cause age-hardening owing to the completion of the A_2 change on tempering at temperatures up to 180 deg. Cent. Other causes of age-hardening effects as well as the three which have been mentioned are, gas reactions within a metal lattice, the transformation on reheating of a single solid solution after quenching into one or two phases of which either or both may be harder than the original phase, and recrystallisation of a cold-worked metal or alloy to a fine grain size or to precipitation of a new phase along the planes of maximum slip.

The development of age-hardening alloys is in a relatively early stage, and of the vast number of possible metal combinations it can be expected that many more alloys will be found which possess age-hardening and other properties which will find useful practical application. In considering methods of field search, it is pointed out that one is to vary the composition and heat treatments of the alloys already known to be age-hardening, and to add other elements. In work on high speed steels, for example, cobalt and vanadium have been added and molybdenum substituted for tungsten. An interesting example of substitution is shown by an alloy containing 50 per cent. nickel, 36 per cent. cobalt, 8 per cent. molybdenum, 5 per cent. chromium, and 1 per cent. carbon. As cast the Brinell hardness was 192, which increased to a maximum of 235 on reheating to 500 deg. Cent. When iron was substituted for the nickel, the corresponding hardness values were 500 and 817. The second method of field search is to make a study of promising binary or ternary alloys rapidly in a qualitative manner, and in ternary systems this is most advantageously done by maintaining the ratio of two elements constant and varying the third.

Alloy.	Hardness as quenched	Hardness after tempering at deg. Cent.								Maximum hardness increase.
		200.	300.	400.	500.	600.	700.	800.	900.	
98 Ni, 2 Be	140	150	165	165	375	311	262	202	145	235
95 Ni, 5 Be	302	302	322	340	480	480	340	286	296	178
10 Co, 88 Ni, 2 Be	185	185	187	192	293	387	269	217	179	202
99·6 Cu, 0·4 Be	50	53	50	42	42	38	34	47	42	3
97 Cu, 2·6 Co, 0·4 Be	69	69	74	80	143	200	92	80	80	131
85 Fe, 15 W	147	145	147	149	147	156	187	230	202	83
15 Co, 73 Fe, 12 W	286	277	286	293	321	351	435	293	228	149
97 Fe, 3 Be	212	223	248	277	364	418	321	277	277	206
15 Co, 82·5 Fe, 2·5 Be	255	255	277	320	387	460	387	340	302	205
52 Ni, 36 Co, 9 Fe, 3 Ti	150			150	163	207	286	185	143	136

It has been found that the addition of cobalt to several alloys has the effect of markedly increasing the extent to which they age-harden. The results obtained with alloys of a number of different types are included in the accompanying table, where the figures given in the final column indicate the difference in hardness between the alloys as quenched and after heating to the temperature yielding the greatest hardness increase. The nickel-beryllium alloys exhibit marked age-hardening properties and resistance to oxidation at high temperatures. The effect of adding cobalt to the alloy containing 2 per cent. beryllium is to raise the temperature of maximum hardening by 100 deg. Cent., but actually the amount of hardening which occurs is slightly less. The copper alloy containing 0·4 per cent. beryllium reaches its maximum age-hardening of 53 by heating at 200 deg. Cent., and this is raised to 200 after ageing at 600 deg. Cent. by the addition of 2·6 per cent. cobalt. The addition of cobalt also has the effect of raising the electrical conductivity of the alloy to 45 per cent. from the value of 38 for the alloy not containing cobalt. At room temperature the alloy containing 0·4 per cent. beryllium and 2·6 per cent. cobalt has an elastic limit of 61,000 lb. per square inch and a tensile strength of 90,200 lb. per square inch, and at 350 deg. Cent. the corresponding values are 61,000

and 67,500. The addition of 15 per cent. cobalt to the 85/15 iron tungsten alloy considerably increases the hardening obtained, and tends to lower the temperature of maximum hardening, but with this exception the effect of cobalt is to raise the temperature. At 800 deg. Cent. the elastic limit, tensile strength, and elongation of the straight 85/15 alloy are 28,000 lb. and 32,000 lb. per square inch and 31 per cent., while the corresponding values on the alloy containing 15 per cent. cobalt are 83,000 lb. and 107,000 lb. per square inch and 3 per cent. The effect of adding cobalt to iron-beryllium alloys is to raise the intrinsic hardness of the alloy, but there is no appreciable increase in the amount of hardening on ageing.

Torsion Impact Tests.

THERE is a great need for some useful method of assessing the shock-resisting properties of different hard steels. The usual notched bar impact tests are useless for the purpose, as all very hard materials give such consistently low results that there is not a sufficient possible variation for any discrimination to be made between different specimens; and yet many extremely hard materials in the form of tool steels or rock drills as well as various high-carbon (but not specially hardened) steels, such as rails, are known not only to possess a serviceable resistance to shock, in spite of their low notched bar figure, but to show considerable variation in shock resistance for a given hardness value.

To deal with this class of material, G. V. Luerssen

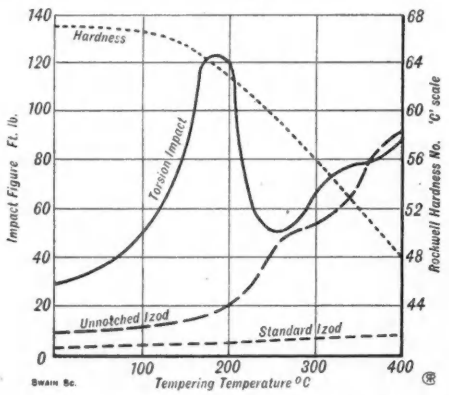


FIG. 1—*Relation between Izod and Torsion Impact Tests on 1·06% carbon steel, brine quenched and tempered (Luerssen and Greene).*

and O. V. Greene* have recently described a machine for carrying out torsion impact tests and have used it to explore the possibility of detecting variations in the impact resistance of hardened steels.

They indicate that in the usual notched bar test, and even in the bending of unnotched bars by impact, the stress concentration is too great in very hard materials, and they therefore make use of a torsion test in which the suddenly applied stress can be distributed over a cylindrical length of specimen. The test piece is made to fit at one end into a sleeve which may be moved axially in a rigid housing and at the other end carries a cross arm. A fly-wheel mounted axially with the test piece and carrying two symmetrical bosses is rotated by means of a handle (or motor) and brought up to a predetermined speed. A knob is then pushed sharply, causing the test piece to slide inwards and bringing the cross arm into engagement with the striking bosses of the rotating wheel. The specimen is thus broken under torsional impact. The residual speed of the wheel is then noted and the energy absorbed in breaking the specimen obtained from a table prepared from the previously determined values of the kinetic energy corresponding to various wheel speeds. The general principle of the machine will at once be clear to those who are familiar with the working of the Guillery machine for testing notched bars. The operation of mounting and breaking the specimen is rapidly carried out. It is essential

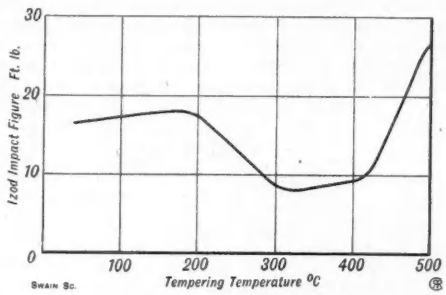


FIG. 2—*Izod Impact Figure of Nickel-Chromium-Molybdenum Steel, showing low values obtained on tempering between 200° and 450° C. (from an Armstrong-Whitworth publication).*

to use a guard over the middle part of the machine during testing, to prevent the flying of fragments, which is said frequently to occur in testing quenched, untempered steels.

To give results of an easily measurable magnitude the dimensions of test piece chosen were 0·25in. diameter over 1in. parallel with 0·5in. square ends, the intermediate contour being designed to prevent stress concentration outside the acting length. Wide variations in initial speed were found to exert a negligible effect, provided that the residual energy of the fly-wheel after fracture was not less than one-quarter of its initial energy. Reproducibility, or the average divergence of duplicate results, was similar to that shown by the Izod test in softer materials.

Detailed tests are reported on three classes of material, 1 per cent. carbon steel, 1 per cent. carbon, 1·5 per cent. chromium steel, and a high-speed tool steel in the first paper quoted, and three carbon tool steels in the second. The tests on 1 per cent. carbon steel are illustrated in Fig. 1 and reveal characteristics which would be unsuspected from the results of notched or unnotched impact bend tests. The difference between the curve obtained with the torsion impact machine and with the Izod machine is very striking; in particular the torsion curve shows a maximum at 175 deg. Cent., which is missing in the Izod test.

Tests on a much wider variety of steels will be necessary before the practical significance of the results of the torsion impact test can be properly

* G. V. Luerssen and O. V. Greene: "The Torsion Impact Test." Proc. Am. Soc. Testing Materials, 1933, 33 pt. 2, 313-327, and "The Torsion Impact Properties of Hardened Carbon Tool Steel." Trans. Am. Soc. Steel Treating, April 1934, 311-337.

assessed and its relation to other impact tests defined. The authors have made a beginning in applying the test to tool steels, though some of their deductions were challenged in the discussion. Meanwhile, however, it is interesting to recollect that in steels with a higher Izod impact figure in the hardened condition there is a very slight increase in its value with increasing tempering temperature up to about 200 deg. Cent., and then a fall with a region of low impact value between 200 deg. and 450 deg. Cent.—Fig. 2. The dip succeeding the maximum in the torsion impact curve occurs over about the same range of temperature as the low notched bar values. On the other hand, the notched bar figure does not show such a considerable rise to a maximum when the tempering temperature is raised to 175 deg., as is shown by the torsion impact figure. It was suggested by Mr. A. V. de Forest that the very low torsion impact figure of the quenched material, which shatters and flies to pieces under the test, was due to internal stresses, and that the rise accompanied the relief of such stresses. Notched bar tests are not affected by internal stress on account of the strictly local concentration of stress which causes fracture, and consequently the notched bar figure is not so likely to be depressed by internal stress in the material as the torsion impact figure.

Zinc Die Casting Alloys.

ZINC base alloys, for a variety of reasons, are eminently suitable for the manufacture of die castings, but many of the earliest castings made were unsatisfactory because of their dimensional instability. To a large extent this difficulty has been overcome by the use of pure metals in making up the alloys, and by careful attention to the composition of the alloys. This dimensional instability is due chiefly to intercrystalline oxidation and the spontaneous decomposition of the beta phase in the zinc-aluminium series corresponding to the compound Al_2Zn_3, into the alpha and gamma phases. The production of reasonably stable zinc base die castings has been rendered possible not only by the exclusion of impurities from the alloys, but also by limiting the quantity of aluminium present, and an alloy which has been standardised by the American Society for Testing Materials, and which, because of its excellent properties, is widely used, has the following specified composition :—Copper, 2·5 to 3·5 per cent.; aluminium, 3·5 to 4·5 per cent.; magnesium, 0·02 to 0·12 per cent., and maximum contents of 0·1, 0·01, 0·005, and 0·005 per cent. for lead, iron, cadmium, and tin respectively, the remainder zinc. With this composition only a small amount of the beta phase is present, and the presence of magnesium has the effect of helping to stabilise the alloy. The transformation of the beta phase of zinc-aluminium alloys is a subject thus not only of considerable scientific interest, but of practical importance. It has been investigated on several occasions, and a further study has been made of it by R. G. Kennedy, of the U.S. Bureau of Standards, and is reported in the May and June issues of *Metals and Alloys*. The investigation consisted of an examination of the properties of the beta phase and methods of retarding its decomposition. In order that the characteristic behaviour of the beta phase should not be obscured by the effect of impurities, the purest aluminium and zinc obtainable were used. The former metal contained 0·02 per cent., and the latter less than 0·0001 per cent. of total impurities. Two alloys containing 85 and 78·3 per cent., referred to respectively as A and Z, were prepared, and after casting were subjected to a homogenising treatment, which consisted of heating them for thirteen days at 350 deg. Cent.

Measurements of heat evolution after quenching from 350 deg. Cent. showed that its intensity was proportional to the relative amount of beta present. Maximum evolution occurred about 4½ minutes after quenching, in the alloy consisting wholly of beta, and in 1½ minutes in the alloy consisting of alpha and beta. Hardness determinations with a Brinell machine using a $\frac{1}{16}$in. diameter ball and a deadweight load of 12·6 kilos., on specimens quenched in ice water from 350 deg. Cent., showed that on standing at 24 deg. Cent. after quenching, alloy Z increased in hardness to a maximum value in six minutes, and alloy A in two minutes. Both alloys decreased in hardness after the maximum had been reached, the hardness curves being similar to the heat evolution curves. The hardening is associated with the decomposition of the beta phase with the formation of an extremely fine eutectoid structure of the alpha and gamma solid solutions. It is perhaps to be accounted for by the straining of the lattice involved in this transformation, and relieving of the lattice strain probably explains the subsequent decrease in hardness, since no microscopic evidence has been obtained of the coalescence of fine grains to large ones. X-ray examination has shown that the beta phase is a simple solid solution above 256 deg. Cent., and has a face-centred cubic structure similar to that of aluminium. According to dilatometric measurements on a quenched alloy containing 77·6 per cent. zinc, most of the contraction at 24 deg. Cent. occurred during the first ten minutes after quenching, although a slight change continued over a period of about four days.

When alloys A and Z were kept in ice water after quenching, the maximum hardening was not reached until after 15 and 35 minutes respectively, instead of 1½ and 6 minutes at 24 deg. Cent. Alloy Z maintained at −70 deg. Cent. did not appreciably change in hardness over a period of seventy minutes, which suggests that at this temperature the decomposition of beta is suppressed. Confirmation of this retarding effect at low ageing temperatures was obtained from further dilatometric experiments with the alloy containing 77·6 per cent. zinc, which showed no contraction, save that attributable to change in temperature, occurred at −190 deg. Cent., although when the specimen was allowed to return to room temperature the normal contraction consequent upon the beta change occurred. The curve showing the effect of ageing temperature on the transformation rate indicates that at about 250 deg. Cent. the decomposition would take place practically instantaneously, and that below about −20 deg. Cent. no decomposition would occur.

It has already been demonstrated by Pierce and Brauer that some metals, including copper and magnesium, when added to zinc-aluminium alloys in the beta range of composition, have the effect of retarding the decomposition of the phase, and further experiments were made by Kennedy in the course of his investigations, on the influence of magnesium. The addition of 0·21 per cent. magnesium reduced the shrinkage after quenching to a very small amount, but X-ray evidence indicates that the decomposition

of the beta phase was not suppressed even with 0·4 per cent. magnesium, from which it is inferred that the shrinkage which accompanies the change has been balanced by an expansion resulting from the precipitation of some magnesium compound. Some support is given to this conclusion by the observation that the hardness increase which occurred in an alloy containing magnesium was permanent.

A paper of a more practical nature on the same subject by E. A. Anderson and G. L. Werley appears in the May issue of *Metals and Alloys*, and deals with the testing of modern zinc base die casting alloys and the relation of such tests to the practical application of the materials. To be of much value tests of this kind should indicate whether the two factors giving rise to dimensional instability and change in properties, namely, intergranular corrosion or oxidation, and internal phase changes, are likely to be present under a given set of conditions. The tests were carried out on alloys containing 4·1 per cent. aluminium, copper varying from 0 to 2·9 per cent., and magnesium varying from 0 to 0·04 per cent. After casting, such alloys shrink on standing from a little less than 0·004in. to a little more than 0·005in., according to the composition, on a 6in. length. Most of the shrinkage occurs within the first five weeks, and thereafter over a period of six months is almost immeasurably small. Over periods of about three years some of the alloys showed an expansion. At 95 deg. Cent. the contraction which required five weeks at room temperature occurred in about three hours, and practical use may be made of this fact by subjecting the castings to a stabilising anneal at this temperature before finish machining. The extent of the dimensional changes, subsequent to the initial shrinkage, appears to increase with the copper content. None of the alloys considered are liable to intercrystalline corrosion, because of their composition and purity, but if contaminated by impurities they become susceptible to it in warm, humid atmospheres. Susceptibility to intercrystalline corrosion is determined by the steam test, which consists in exposing the alloy at 95 deg. Cent. to an atmosphere saturated with moisture. In order to ascertain that the composition is satisfactory in this connection, the test has been adopted by the A.S.T.M., and according to the requirements the material must stand up to the steam test for ten days without showing any intercrystalline corrosion.

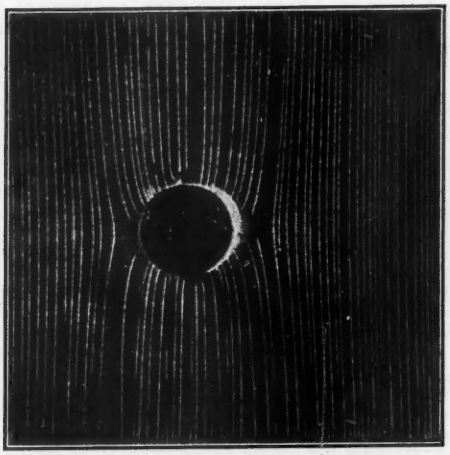
Changes in the mechanical properties accompany the dimensional changes, and considerable decreases in impact values were observed in some of the alloys, those containing copper showing the most appreciable losses. Some of the alloys of relatively low copper content did not show much change in impact strength after three years' indoor ageing, but showed marked loss at 95 deg. Cent. The alloy containing no copper retained its impact strength well when exposed either at room temperature or 95 deg. Cent. under dry conditions for prolonged periods. Although tensile strength is not a property of much consequence in most of the applications in which zinc base die castings are used, the results of ageing tests are of interest in showing that on indoor ageing for periods up to three years the change, which in most cases was a decrease, is comparatively small. At 95 deg. Cent., however, even the alloy which showed an actual increase at room temperature suffered a loss in tensile strength. Elongation values on alloys of this class are low, and to the relatively small changes which have been observed little or no significance can be attached.

The Distribution of Elastic Strain in Metal Specimens.

WHILE photo-elasticity, or the use of polarised light, affords a means of studying elastic deformation in transparent materials, it is in general only the permanent deformation of the surface of a metallic specimen which can be detected by methods which have hitherto been proposed—such as the direct observation of the appearance of Lüders' lines, etching with Fry's reagent, inducing reactions which give rise to coloured products within a thin layer of gelatine in contact with the surface, detailed exploration of hardness, or direct measurement of the deformation of a network of lines inscribed on the surface.

Since, in design, it is the elastic deformation which has to be considered, an accurate knowledge of the distribution of stress before the elastic limit is reached is important.

A method of studying elastic deformation in metal specimens is described by A. Portevin and M.



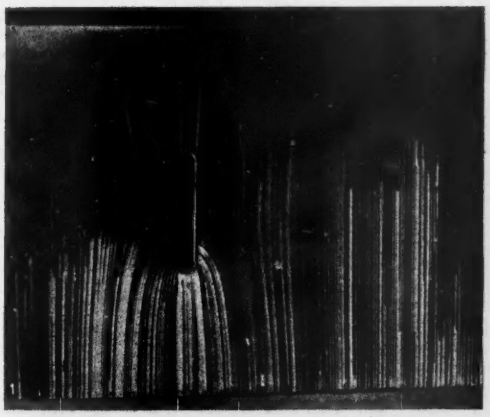
BENDING: SINGLE ROUND HOLE

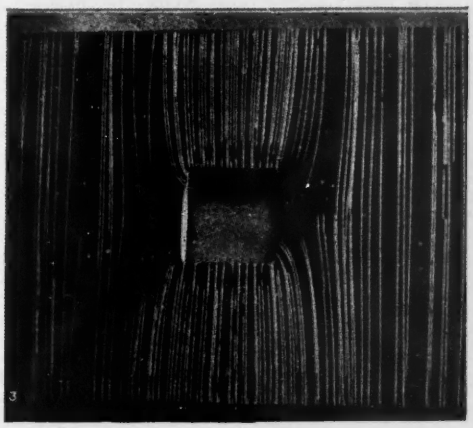
Cymboliste in the *Revue de Métallurgie*, April, 1934 (Vol. 31, pages 147 to 158), in amplification of a note presented to the Académie des Sciences on August 7th, 1933. The test piece or specimen is covered with an adherent varnish which is sufficiently plastic to adapt itself to the initial deformation of the metal, but sufficiently fragile to break before the deformation ceases to be elastic. Very fine fissures or cracks, visible under appropriate illumination, are thus produced. They appear in regions of highest tensile stress, but are not produced in regions of compression. They are held to indicate the points at which fatigue failure is liable to originate, and to trace out the path which fatigue cracks would follow in the metal.

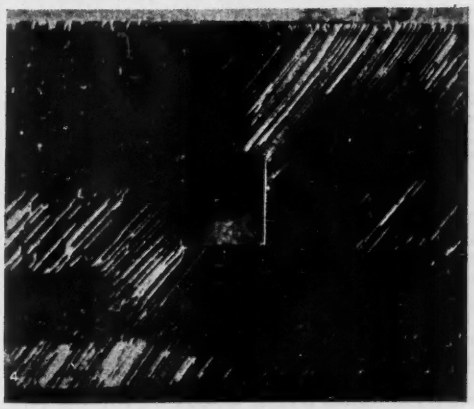
The degree of adhesion, and the properties of the varnish depend on the preparation of the surface, the composition of the varnish, and the conditions of drying. Portevin and Cymboliste made a number of trials with varnishes in which the nature and proportions of the constituents were varied. The materials they experimented with included :

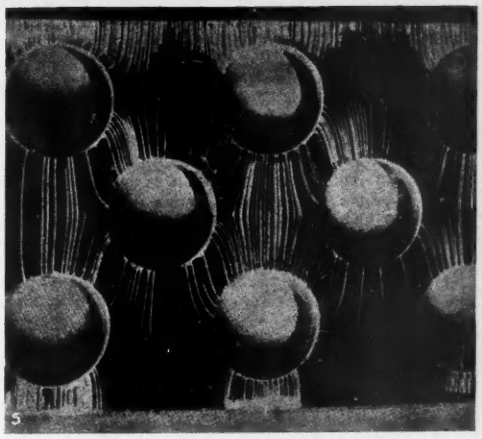
(1) Gums and resins, both natural and artificial,

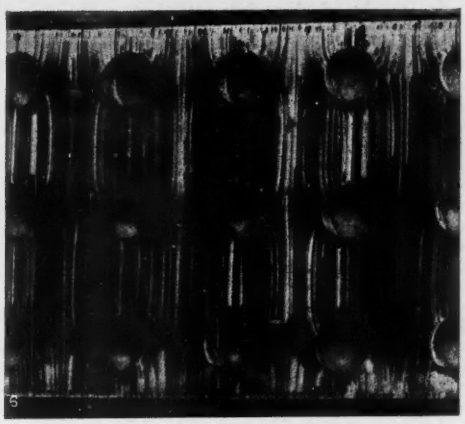
ELASTIC STRAIN IN METAL SPECIMENS











1. BENDING: THIN TRANSVERSE SLOT
2. BENDING: OBLIQUE SLOT
3. BENDING: SQUARE HOLE
4. TORSION: SQUARE HOLE
5. BENDING: ROUND HOLES, DIAGONAL PITCH
6. BENDING: ROUND HOLES, SQUARE PITCH

such as copal from various sources, and condensation products of the phenols (bakelite, colophane, &c.), alone or mixed.

(2) Solvents; various alcohols, esters (amyl-acetate, &c.), aromatic hydrocarbons (toluene, &c.), alone or mixed.

They point out that the properties required approximate to those of a commercial varnish of inferior quality. The product must have a tendency to crack, without doing so spontaneously during drying. The composition of the varnish ought, therefore, to be chosen in relation to the temperature of the test and the sensitivity desired—that is, the value of the deformation at which it is intended that the fissures should appear. The mechanical properties of the varnish should be modified to suit, among other things, the amplitude of the elastic deformation and the speed of straining. The lower the modulus and the more rapid the deformation, the more important is the plasticity of the varnish, though it is noted that a high degree of plasticity has the disadvantage of causing the fissures to join up and disappear in process of time. For metals of high elastic modulus such as steel, tested statically, it is necessary to have a very fragile varnish without sensible viscosity other than that which permits of drying without spontaneous formation of fissures.

From these trials a certain number of more or less satisfactory formulæ have been arrived at for varnishes which dry spontaneously in air or on baking at 100 deg. to 150 deg. Cent. Varnishes which require baking at a higher temperature should be avoided. Suitable properties can be obtained from a variety of mixtures.

The paper contains more than thirty photographs illustrating the action of tension, torsion, and bending on plain and on notched bars, and showing the effect on distribution of tensile stress of narrow slots set at various angles, of square and round holes, and of several holes differently arranged in plates subjected to tension, torsion, or bending. A few of these are reproduced herewith. By suitable choice of varnish it is claimed that the method may be adapted for microscopical examination, though with rather less regular results, to show the effect of very small defects or discontinuities, like non-metallic inclusions in steel or graphite in cast iron.

The authors refer to two previous instances of the use of similar methods. O. Dietrich and E. Lehr (Zeit. V.d.I., 1932, 76, 973–982) described a method of determining the magnitude and direction of stresses in various machine parts by the use of a special varnish, the composition of which was not stated. They investigated the influence of shape of the stressed part and discussed the bearing of their results on design. The Maybach Works (*Engineering Progress*, 1932, 13, 264–267) developed a similar method for determining the distribution of stresses in machine parts subject to alternating loads, by coating the part with a lacquer which would fissure but not flake off under distortion. They, also, gave no information as to the conditions of test or the composition of the lacquer used. Portevin and Cymboliste have, unfortunately, followed the example of these previous investigators in giving no indication, except in the general terms already quoted, of the composition of the varnish which they found most suitable, and in not even giving the composition used on the specimens which provide the illustrations of their paper. They apparently prefer that the detailed specification of a suitable lacquer, based on the information which they give, should be left to those who desire to adopt this useful method for determining the distribution of stresses arising in machine parts from certain conditions of design or use. It is of interest to note that they refer also to the fact that work is in progress on a method of optical examination of transparent elastic coatings applied to metal specimens with the same object in view.

Nickel-Silver-Copper Alloys.

THE number of alloy systems which are known to include alloys which exhibit age-hardening phenomena is now very large, and appears to be still rapidly increasing; in consequence, it is not customary to regard the discovery of one additional example of an age-hardenable alloy as outstandingly important unless the new alloy has a combination of properties which are sufficiently remarkable to indicate possibilities of useful commercial applications. The alloys of copper with nickel and silver which have recently been studied by Pfister and Wiest (*Metallwirtschaft*, XIII, Nr. 18, s. 317) seem

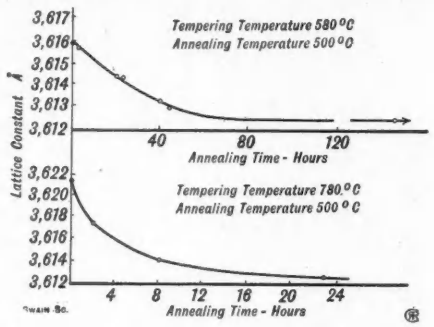


FIG. 1—*Kinetic Curves of Lattice Constants for Cu+1% Ni+4% Ag Alloy at Varying Supersaturations.*

worthy of consideration both from the commercial and scientific points of view.

Prior to the work of Pfister and Wiest the constitution of these alloys had been studied by Guertler and Bergmann (*Z. Metallkunde* 25, 1933, 53), who point out that the silver-rich phase which occurs in the two-phase field adjacent to the copper-rich phase field is silver containing some copper, but practically no nickel in solid solution. Hence, if a copper-rich ternary nickel-silver-copper alloy is slowly cooled from a temperature at which it had a homogeneous constitution any particles of a second phase which separate are nickel-free and of virtually the same composition as the particles which would separate under similar conditions from a binary silver-copper alloy. Similarly, it may be argued that the phases separated during ageing of supersaturated ternary and binary alloys are identical, and thus it is to be anticipated that the ageing behaviour of the nickel-bearing alloys would run parallel to that of the nickel-free alloys.

The ageing of binary silver-copper alloys has been investigated by Ageew, Hansen, and Sachs (*Z. Physik*, 1930, 66, 350), by Smith and Lindlief (Amer. Inst. Min. Met. Eng., "Tech. Pub.," No. 433, 1931), and by Wiest (*Z. Physik*, 1932, 74, 225). From these

researches it may be stated that in normal polycrystalline aggregates no age hardening occurs, but that precipitation of a second constituent takes place. This precipitation is indicated by decrease in the lattice parameter of the solid solution phase, but the change in parameter does not occur progressively as a steady decrease, evidence being afforded by the X-ray spectra that at any one time lattices of smaller and larger parameter than the average are present. In single crystals of binary silver-copper alloys, on the other hand, Wiest has observed that the solubility of silver is greater than in the polycrystalline alloys, and the parameter change during ageing is a steady decrease. These single crystals were prepared from the molten alloy and have a pronounced mosaic structure.

Pfister and Wiest's study of the ternary alloys commenced with the determination by X-ray methods of the solid solubility of silver in copper containing 1 per cent. of nickel by weight, the alloys being in the rolled and annealed (polycrystalline) condition. The results are summarised in the table below, which also includes Wiest's figures for the solubility of silver in nickel-free copper polycrystalline aggregates. It will be seen that the presence of nickel lowers the solubility of silver, both at temperatures above 600 deg. Cent. and temperatures below 450 deg. Cent.:

Temperature, deg. Cent.	Silver dissolved, weight per cent.	
	In copper.	In 1% Ni alloys.
370	0·5	0·3
450	0·9	0·9
500	1·4	1·3
600	2·6	2·6
700	4·7	4·4
770	7·4	6·3

Pfister and Wiest then investigated the ageing behaviour of ternary alloys containing 1 per cent. nickel, together with 4, 5, and 7·5 per cent. silver, by determining the lattice parameter and Brinell hardness of alloys at room temperature after they had been cold rolled, quenched from a variety of temperatures, and further reheated for a series of periods at various tempering temperatures. The X-ray photographs indicated that in these alloys, as in the cast single-crystals of binary silver-copper alloys, there is a smooth decrease in the lattice parameter during ageing of quenched material. This result they interpret as being due to the influence of mosaic structure upon the ageing, as the mosaic structure is considered to be well developed in these ternary alloys. The X-ray work indicates the manner in which the approach of the alloys to a state of equilibrium at the tempering temperature is influenced by such factors as composition, quenching temperature, and tempering temperature. In Fig. 1 are reproduced the curves connecting lattice constant with duration of annealing at 500 deg. Cent. of two specimens of the alloy containing 4 per cent. of silver and 1 per cent. of nickel quenched in the one case from 580 deg. Cent., and in the other from 780 deg. Cent. The curves become asymptotic at the same value of the parameter, but after markedly different periods of annealing—about twenty-four hours in the one instance and eighty hours in the other. In alloys containing either 5 or 7·5 per cent. of silver, together with 1 per cent. of nickel quenched from 780 deg. Cent. or 700 deg. Cent. respectively, the parameter assumes its asymptotic value after but six hours at 500 deg. Cent.; hence it may be concluded that increase in quenching temperature, or in concentration of silver, markedly accelerates the attainment of equilibrium by these alloys. In passing, it may be pointed out that the tracing of these "kinetic curves," as they are termed by Pfister and Wiest, from the initial to the final value of the lattice constant, is a method of investigation which should increase our knowledge of the kinetic reactions in the solid state, in addition to constituting the most reliable method of determining the equilibrium value of the parameter, and hence the solubility lines of the phase diagram.

The study of the ageing by means of Brinell hardness measurements included experiments upon the three alloys employed in the X-ray investigation, but results are given only for the alloy containing 5 per cent. of silver and 1 per cent. of nickel. The hardness values obtained by annealing this alloy at 335 deg. Cent. after quenching from 780 deg. Cent. are plotted in Fig. 2. It will be seen that the curve drawn through the experimental points passes

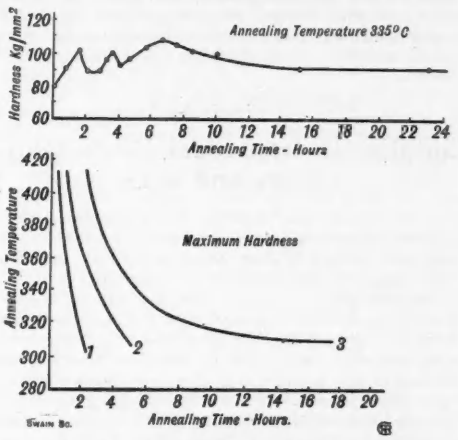


FIG. 2 (*Above*)—*Curves of Hardness for Alloy Cu+5% Ag+1% Ni.*

FIG. 3 (*Below*)—*Curves of Critical Dispersion for Alloy Cu+5% Ag+1% Ni.*

through three successive maxima. This type of curve has apparently been obtained at a variety of temperatures within the range 310 deg. to 410 deg. Cent., since the authors draw three smooth curves upon a graph of annealing temperature against duration of treatment—see Fig. 3—to indicate the attainment of maximum hardness. These curves, following the example of Meissner (*Z. Metallkunde*, 17, 1925, 77), they term "curves of critical dispersion." As the observation of three successive hardness maxima in this instance appears to be of considerable importance from a theoretical standpoint, it is to be regretted that the authors have not set out *in extenso* the data upon which the curves are based. In this connection it would be particularly interesting to know on the average of how many individual hardness measurements each of the points included in Fig. 2 is based. Whilst it has been suggested by Herbert (*Journal*, Inst. Metals, 48, 1932, 211) that age-hardening curves consist of a sequence of periodic fluctuations, it appears to be the more general opinion that when a single phase is precipitated during the ageing of an alloy the true curve relating hardness to time is smooth and passes through or approaches a single maximum. Fluctuations about a mean curve which may result in the appearance of more than one maximum are usually attributable to

the number of hardness determinations being insufficiently large to average out experimental errors and point-to-point variations in hardness.

Although these ternary nickel-silver-copper alloys do not harden greatly during ageing, the increase observed by Pfister and Wiest being about 20 Brinell, the rate at which the age hardening occurs is remarkably rapid when compared with hardenable copper alloys in general. For example, Corson's nickel-silicon-copper alloys require some days to reach maximum hardness at 350 deg. Cent. as against a few hours at 335 deg. Cent. in this instance. Here, then, we have a series of copper alloys which will age harden at comparatively low temperatures, but temperatures at which tough pitch or arsenical coppers have a limited life under continuously applied stresses. As strengthening due to age hardening may offset softening due to recrystallisation, copper containing nickel and silver appears to be a material worthy of consideration for service under continued load at such temperatures as 300 deg. Cent., as, for example, stay rods in locomotive fire-boxes.

Sulphur in Iron, Steel, Ferro-Alloys, Slags and Ores.

THE combustion method for the estimation of sulphur in iron and steel has been known for a considerable time. It has been improved especially with regard to the apparatus by Holthaus—*Stahl u. Eisen*, 44 (1924), pages 1514/19. The sample is burnt in a stream of oxygen and the sulphur dioxide formed is absorbed in an alkaline hydrogen peroxide solution, and the excess of alkali is titrated with sulphuric acid using Congo Red, or alizarinsulphonic acid, as indicator. The method yields good results, but the detection of the end point is very difficult as mixed colours occur. On account of this difficulty, Thanheiser and Dickens—*Arch. Eisenhuttenwes*, April 7th, 1934, pages 557/62—have investigated the possibility of employing a potentiometric method for the titration. In an earlier investigation—"Mitt. K.W. Inst. Eisenforsch.," 12 (1930), pages 203/23, and "Chem. Fabrik," 4 (1931), pages 145/47—they found that the hydrogen, quinhydrone, oxygen or oxide electrode in combination with a normal calomel electrode, can be utilised for acid-alkali solutions of $^1/_{10}$, $\frac{1}{2}$ and normal strength.

In further experiments they used an oxygen electrode for $^1/_{20}$, $^1/_{40}$, and $^1/_{100}$ normal solutions, but considered the plotting of a graph wearisome and not suitable for a quick method. For this reason they chose an end point (*umschlags*) electrode. The potential of this type of electrode at the equivalence point is equal to that of a platinum electrode immersed in the titration vessel. It is essential, however, that the potential of such an electrode should remain constant, and the authors, therefore, used that recommended by Laur—"Dissert. Tech. Hochsch.," Dresden, 1929—$Hg/Hg_2(CH_3COO)_2$, $2n-CH_3COO\ N\ a$.

The absorption vessel of the Holthaus apparatus was modified for use with this electrode. A platinum electrode was fused in and a side opening made into which was fitted, by means of a rubber stopper, the connecting tube of the electrode vessel. The connecting tube was closed at each end with a suitable diaphragm, and contained a saturated solution of potassium sulphate. A platinum wire was fused through the bottom of the electrode vessel and covered with mercury. Solid mercurous acetate was placed on top of the mercury, and then a twice normal solution of sodium acetate. A sensitive galvanometer or an ordinary millivoltmeter was used as current indicator. The titration was carried out in the usual way, except that the end point was indicated by the null point of the galvanometer instead of by the colour change of an indicator. Good results were obtained by this method on samples of carbon steel, alloy steel, cast iron, and alloy cast iron.

In order still further to simplify the method, experiments were conducted on a direct titration with dilute caustic soda solution of the sulphuric acid formed in the absorption vessel. It was necessary to determine whether the sulphur dioxide is quantitatively absorbed and oxidised in neutral or weakly acid hydrogen peroxide solution, and if any influence is exerted by the carbon dioxide produced from the carbon contained in the sample.

Experiments with solutions containing known quantities of sulphur dioxide to which were added varying concentrations of hydrogen peroxide solution indicated that 50 c.c. of a $\frac{1}{2}$ per cent. solution of hydrogen peroxide sufficed for the quantitative oxidation of 50 mgm. of sulphur.

A stream of carbon dioxide was passed into a potentiometrically neutralised solution of hydrogen peroxide and an absorption of carbon dioxide was indicated by a deflection of the galvanometer. It was also noted that the absorption was greater with higher concentration of hydrogen peroxide. However, a blank estimation carried out by the combustion of 0·1 gramme of acetylene black indicated that only a small amount of carbon dioxide was absorbed, and this was completely expelled again by prolonging the oxygen stream for three minutes after the end of the combustion. This indicates that the sulphur estimation is not detrimentally influenced by the quantity of carbon dioxide produced during the combustion of a steel.

Experiments with standard caustic soda solutions indicated that the addition of one drop (0·03 c.c.) of N/100 solution caused a deflection of two divisions on a galvanometer whose sensitivity was of the order of $0{\cdot}6 \times 10^{-6}$ ampère for each division.

As a result of the experiments described above, the estimation of sulphur in a steel is carried out in the following manner:—200 c.c. of perhydrol and 1 c.c. of sulphuric acid (1 : 6) are made up to 10 litres with water. The sulphuric acid is added in order to increase the conductivity and also the keeping qualities of the absorption solution. The blank on this solution is determined by adding about 50 c.c. of water to 50 c.c. of the solution, and titrating with caustic soda solution whilst maintaining a constant stream of oxygen. The titer of the caustic soda solution is checked by the combustion in excess of oxygen of 2 grammes of a standard steel sample at 1150 deg. to 1200 deg. The oxygen stream is continued for three or four minutes after the end of the combustion. Rubber tubing is to be avoided as far as possible in the construction of the apparatus. Good results are obtained by this simplified method.

Vita—*Stahl u. Eisen*, 40 (1920), pages 933/35—has investigated the combustion method for the estimation of sulphur in slags, ores, and fuels. In these materials the sulphur exists not only in the form of sulphides and sulphates of iron and manganese, but may also be present combined with calcium, magnesium, zinc, lead, barium, and the alkali metals. According to Vita, barium sulphate is not decomposed by heating in oxygen, sulphates

of the alkali metals only in small degree, and for the complete decomposition of other sulphur compounds contained in slags the sample must be heated for three-quarters of an hour at 1350 deg.

The authors therefore investigated the possibility of decomposing barium sulphate and the alkali sulphates in order to render the method more generally applicable and independent of the composition of the sample.

Experiments were made with freshly precipitated and ignited barium sulphate, and the titer of the caustic soda solution was adjusted by means of pure anhydrous manganese sulphate, which is decomposed quantitatively at temperatures over 1000 deg. One-tenth gm. of barium sulphate was mixed with various fluxes, such as sodium carbonate, silica, aluminium drillings, ferro-silicon, graphite, lead chromate, and electrolytic iron in varying amounts, and the mixture burnt at 1150 deg. to 1250 deg. The results indicate that the extent of the decomposition depends on the amount of flux and the temperature of combustion. A mixture of graphite and silica, lead chromate, and electrolytic iron are the most promising for the complete decomposition of barium sulphate.

Electrolytic iron is the most suitable on account of the small blank and the high temperature developed during the combustion which favours the decomposition of the barium sulphate. Experiments were, therefore, continued with electrolytic iron using three slag samples containing respectively 0·29, 1·65, and 2·64 per cent. of sulphur, a proportion of which was in the form of barium sulphate. It was found that the best mixture to use was 1 gramme of slag and 2 grammes of iron, and that the iron must be in the finest possible form of drillings. With all three samples, the decomposition increases proportionately with the temperature, and is quantitative at and over 1300 deg. A quantitative decomposition was only obtained when the boat contained a completely homogeneous fusion. Similar results were obtained with potassium and sodium sulphates. Estimation of the sulphur by this method in slags from the blast-furnace, basic and acid open hearth, basic converter and high-frequency furnace, yielded very good values.

The caustic soda solution was standardised by the combustion of a mixture of 0·1 gramme barium sulphate and 2 grammes of iron. A temperature of 1300 deg. to 1350 deg. was employed.

With open-hearth and electric furnace slags containing fluorides, high values were obtained owing to a portion of the fluorine being driven over and titrated. Various expedients were tried in order to surmount this difficulty, and finally titration of the sulphurous acid with iodine solution was adopted, the oxidation with hydrogen peroxide being discontinued. Owing to the influence of the stream of oxygen it was not possible to add excess of iodine solution and titrate back with sodium thiosulphate. The absorption of the sulphur dioxide in water and titration of this solution with iodine also gave low values as the oxygen tended to displace the sulphur dioxide from solution. This difficulty was overcome according to the method recommended by Kassler ("Untersuchungsmethoden für Roheisen, Stahl, und Ferrolegierungen," Stuttgart: Ferdinand Enke, 1932, pages 42/48); 100 c.c. of water and about 10 c.c. of starch solution (5 grammes per litre) were placed in the absorption vessel and a few drops of iodine solution added. As the inflow of the combustion gases tended to decolourise this mixture, first, at the bottom of the vessel, iodine solution was run in as required so as to maintain the initial blue colour. Estimations by this method on pure manganese sulphate, barium sulphate, and slag free from fluorides yielded the correct values.

Estimations were also made on barium sulphate and a slag sample to which calcium fluoride was added in quantities equivalent to from 0·1 to 15 per cent. of fluorine. Correct values within the limits of error were again found.

An additional difficulty with slags containing fluorides was due to the fact that a deposit occurred in the delivery tube which tended to choke the fine jet of the Holthaus apparatus. This delivery tube was, therefore, replaced by a simple glass tube with a 3 mm. bore, and although the gas bubbles were not so finely divided by this arrangement, the results were not adversely affected.

Sulphur can be determined in ores in a similar manner, but there is a danger that the water contained in many ores may, owing to condensation in the delivery tube before reaching the absorption vessel, result in low values.

Determinations were, therefore, made on a mixture of 0·1 gramme barium sulphate with 2 grammes of iron to which was added crystallised barium hydroxide in amounts varying from 0·1 to 1·0 gramme. With up to 0·5 gramme of barium hydroxide (17 per cent. of water) the correct value for the sulphur was obtained when the oxygen stream was continued for three minutes after the end of the combustion period. With 1 gramme of barium hydroxide (25 per cent. of water) the value obtained after three minutes was 93 per cent. of the total sulphur value, and seven minutes were required for the correct result. In contrast to these results experiments in which similar quantities of water were placed in the delivery tube indicated that proportionately longer time was required for the gas flow in order to obtain correct values. This is due to the fact that the condensed water has a higher temperature than water which is merely added and, therefore, absorbs less sulphur dioxide. Although correct values can be obtained in the presence of condensed water, it is advisable to keep the delivery tube as dry as possible so as not to prolong unduly the time required for an estimation.

This application of a potentiometric titration to the Holthaus method is interesting as it has been recognised for some time that the colour change at the end point is very difficult to detect. However, in a works laboratory it may be difficult to maintain electrical apparatus in good condition, and such a method may bring other troubles in its train. Attention may, therefore, be directed to a modification recommended by Rooney (*Analyst*, April, 1934, page 278), in which a lamp of the daylight type (tinted blue) is substituted for the ordinary incandescent type employed with the Holthaus apparatus. This modification is simple and makes a remarkable difference in the detection of the end point.

It has also been pointed out by Rooney that the use of a glass wool filter for retaining iron oxide carried over during the combustion may be a serious disadvantage as glass wool is frequently alkaline and absorbs sulphur dioxide. Iron-oxide fume is usually produced when the combustion takes place under pressure, the pressure being due to the smallness of the jet in the absorption vessel. A larger jet obviates this difficulty and the use of a filter is not essential. The results are not adversely affected by the use of a larger jet, as confirmed by the present authors.

Recent Calorimetric Investigations on Steel.

THE application of thermo-dynamical principles to the elucidation of the theory of the constitution and heat treatment of steel necessitates the accurate determination of the heat changes which accompany the constitutional changes which the metal undergoes. In view of the importance of this method of theoretical approach to the subject, it is somewhat remarkable that a larger collection of trustworthy data is not available. The difficulty of the work no doubt accounts for there being so much uncertainty about accurate values. But recent work at Aachen, at the Kaiser Wilhelm Institute, Düsseldorf, and elsewhere in Germany, involving refined calorimetric

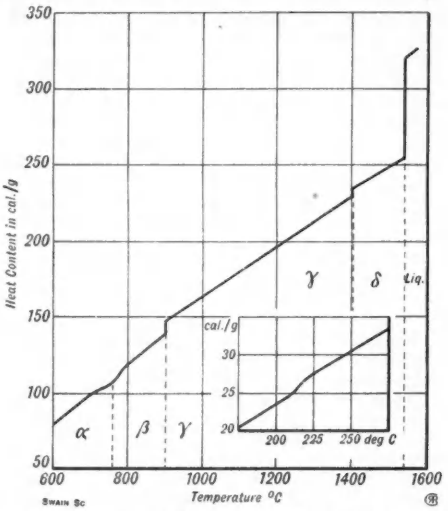


FIG. 1—*Heat content of Pure Iron at Different Temperatures. Inset—Heat content of 1·16 per cent. Carbon Steel between 175° and 275° Cent.*

measurements, will go far in several directions to clear up existing discrepancies. Interest in the subject is reflected by a number of papers, which have appeared in the *Archiv für das Eisenhüttenwesen*. For example, in November last, C. Schwarz[1] made a careful study of all available data having reference to the specific heat and heat content of metals, gases, oxides, slags, &c., which have to be taken into account in making many metallurgical calculations. This formed one of a series of papers by different authors on the heat content of steel and slag, which have an important bearing on equilibrium conditions in open-hearth furnaces. The application of this work is, however, to be found mainly in steel manufacture, while the purpose of this article is to direct attention to researches bearing on constitutional changes in the steel itself.

THERMO-DYNAMICAL DATA FOR IRON AND STEEL.

A valuable critical review of the present state of knowledge of the heat content, specific heat, and specific volume of the iron-carbon alloys has been contributed by Tammann and Bandel.[2] Summarising the work of numerous investigators these authors conclude that the heat content of pure iron shows no discontinuous change up to 906 deg. Cent., though an irregularity appears at A_2 (Fig. 1). At A_3 there is an increase in heat content of 5·7 cal./g.; at A_4 an increase of 1·9 cal./g., and at the melting point a latent heat of fusion of 65 cal./g. The heat change associated with the A_1 point in carbon steels shows a linear increase with increase of carbon content from 0 to a value, varying according to different observers, of 16 to 22·5 cal./g. at 0·9 per cent. carbon, and then a linear decrease as the carbon content rises to 6·67 per cent. (pure Fe_3C). The value adopted by Tammann and Bandel for the A_1 change in 0·9 per cent. carbon steel is 21·2 cal./g. The cementite change A_0 at 215 deg. causes an anomaly in the specific heat of high carbon steels, as shown in Fig. 1.

Numerous determinations of specific volume are discussed. The probable values for pure iron are indicated in Fig. 2. The value adopted for the density at 20 deg. Cent. is 7·875 (specific volume,

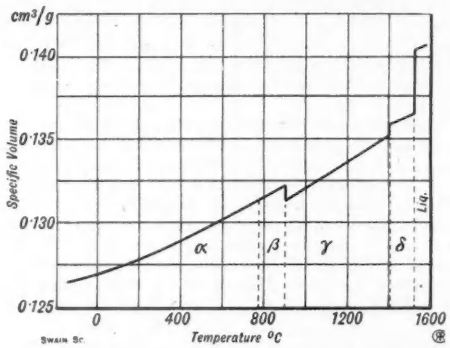


FIG. 2—*Specific Volume of Pure Iron at Different Temperatures.*

0·12699), which, it may be noted, is identical with that obtained by Adcock (7·874), though this value was not mentioned, neither was the still higher (but up to the present unconfirmed) value 7·90 obtained by Tritton and Hanson in 1924. Tables are given in Tammann and Bandel's paper of the specific volume of iron, and of carbon steels at various temperatures. The increase in specific volume (decrease in density) with increase of carbon content was found to be linear at all temperatures until A_1 was reached, being about 0·0006 for each 1 per cent. of carbon at 20 deg., and rather less (0·00054) at 200 deg. Cent. and above. The changes in specific volume associated with the critical ranges are given as :

	Cm.³/g.
A_1 in 0·9 per cent. carbon steel	−0·00040
A_3 in pure iron	−0·00110
A_4 ,, ,, ...	+0·00033
At the melting point of pure iron	+0·0039

Of these figures that which is least certain is the increase in volume on melting, the value adopted being the average of 0·0061 (Desch and Smith), 0·0019 (Benedicks, Ericsson and Ericson), and 0·0039 (Sauerwald and Widawski).

Tammann and Bandel finally discuss, on the basis of the Clausius-Clapeyron equation, the effect of pressure on the temperature of the critical changes in steel, and conclude that under a pressure of 20,000 kilos./cm.² :—

	Deg.	Deg.
The melting point would be raised ..	50 to	1580
A_4 would be raised	136 to	1537
A_3 would be lowered	106 to	800
A_1 in 0·9 per cent. C steel would be lowered	8 to	713

High pressure would thus give a greater range of stability to austenite.

Published data for the heat content of pure iron show discrepancies, especially between 650 and 900 deg. Cent. Esser and Bungardt[3] have noted that the lowest values of the heat content and the highest values of the heat change at A_3 were obtained with ice- or water- calorimeters in which the rate of cooling of the test piece is greatest. The best values of the heat change are, in their opinion, those obtained from measurements of the true specific heat at high temperatures, since, in these, the temperature of the test piece is only slightly altered. After considering all the results summarised in their paper they conclude that the heat content/temperature curve of pure iron below A_2 is convex, and between A_2 and A_3 concave to the temperature axis, and that the best value of the heat of transformation at A_3 is 3·6 cal./g.

HEAT EVOLVED IN TEMPERING QUENCHED CARBON STEELS.

The heat evolved in tempering hardened carbon steels was determined by Wever and Naeser[5] by means of a high-temperature calorimeter, the specimen being placed in the hot calorimeter and the amount of cooling noted. They concluded that a large amount of the austenite-pearlite heat of transformation was liberated during the quenching to martensite, and that the carbon precipitated during the decomposition of austenite at low temperatures can only be combined with alpha-iron to form cementite at higher temperatures.

Stäblein and Jaeger[6] measured the heat liberated in tempering, at 450 deg. Cent., quenched steels with 0 to 1 per cent. of carbon, and found it to increase from 0 to 11 cal./g. In eutectoid steel, it was about 10 cal./g., and of this about 4 cal./g. represents the heat of transformation of the residual austenite, the remainder being heat of precipitation of cementite from the martensitic structure of the quenched steel.

At Aachen much valuable calorimetric work on metals and alloys (*e.g.*, copper, platinum, brass, nickel-silver, &c.), has been carried out under the direction of Hans Esser.[7] More recently, Esser and Bungardt[3] have utilised a specially constructed water calorimeter, to the upper part of which is attached a heating chamber. From this chamber the specimen, brought to the required temperature, could be made to fall into the water of the calorimeter at 15 deg. Cent. By this method they determined the total heat content, at various temperatures from 300 deg. to 900 deg. Cent., first of silver (the results for which they have carefully compared with previous determinations), and then of 20 per cent. manganese steel, and of three carbon steels containing 0·5, 0·8, and 1·4 per cent. of carbon. From the results obtained on the manganese alloy, the heat content of gamma-iron for the range studied, was calculated. The heat change associated with the austenite-martensite transformation in the carbon steels was also estimated, allowance being made for the austenite retained on quenching, and subsequently determined magnetically. In the 0·8 per cent. carbon steel the amount of retained austenite (about 8 per cent.) falls slightly with increasing temperature of quenching above 750 deg. Cent. On the other hand, in the 1·4 per cent. carbon steel, it rapidly rises from 6 to 26 per cent. as the quenching temperature is increased from 750 deg. to 975 deg. Cent. The heat values were deduced from the change in total heat content shown by specimens quenched from just below, and just above, Ac_1 (721 deg. Cent.), and were found to fall from 13·6 cal./g. for the 0·5 per cent. carbon steel to 11·4 cal./g. for the steel with 1·4 per cent. carbon. Taken in conjunction with the data collected by Tammann for the heat change at the austenite-pearlite transformation, it would appear that in the 0·5 per cent. carbon steel the heat of transformation of austenite to martensite is approximately that of austenite to pearlite, but in the 1·4 per cent. carbon steel it is only about half that of the austenite-pearlite transformation.

Esser and Bungardt[4] have continued their work

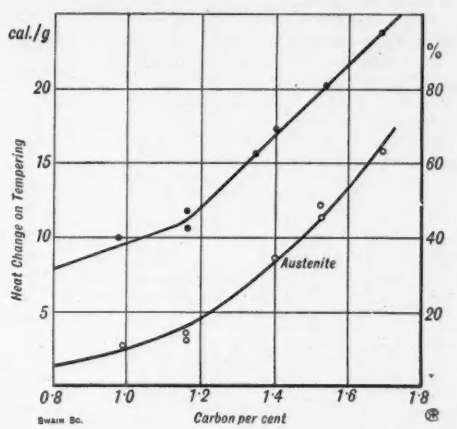


FIG. 3—*Heat evolved on tempering quenched Carbon Steels up to about 600° C., also Austenite Content of the quenched steels.*

with a study of the heat change which occurs on tempering carbon steels containing 1 to 1·7 per cent. of carbon, quenched from the austenite range. By the use of a high-temperature calorimeter, they determined the heat evolved by the quenched steel during tempering at 600 deg. Cent. Combining their results with those of earlier investigators, they found that the heat evolved in tempering the quenched steel up to 600 deg. Cent., increased from 0 to 12 cal./g. as carbon increased to 1·2 per cent., and then more rapidly to 24 cal./g. as carbon rose to 1·7 per cent.

The austenite in the quenched steel determined by magnetic methods increased from 0 to 10 per cent. with rise of carbon to 1 per cent., and then rapidly to 65 per cent. when the carbon reached 1·7 per cent. These results are indicated in Fig. 3.

THE HEAT OF FORMATION OF IRON CARBIDE Fe_3C.

A somewhat different type of investigation is that of G. Naeser[8] on the heat of formation of iron carbide. A review of previous determinations showed large discrepancies between the observed values, extending, at room temperatures, from +8·9 to −15·1 K cal./mol. This variation is mainly due to the fact that direct methods of measurement are not possible, but that the value depends on a small difference between large numbers. The methods used by Naeser and the results obtained were as follows :—

Solution in iodine-potassium iodide	+8·21 K cal./mol.
Solution in copper ammonium chloride	+7·88 „ „

These are in fairly good agreement with E. D. Campbell's value, 8·5 . K cal./mol.

The method of calculation from the heat of combustion of iron carbide, iron, and carbon, respectively,

gives variable results according to the value taken for the heat of combustion of carbon. With finely divided amorphous carbon, Naeser obtained a heat of combustion of 107·86 K cal./mol, which would give +9·6 K cal./mol. for the heat of formation of Fe_3C; but preheating the carbon in order to graphitise it lowered the heat of combustion to 97·4, and this suggests that the value previously obtained by Roth for the heat of combustion of β-graphite, viz., 94·3, may be correct. Assuming this value, the heat of formation of iron carbide is calculated to be −3·9 K cal./mol.

A third method employed was the decomposition of iron carbide by sulphur with the formation of iron sulphide and carbon. This gave a value +1·1 K cal./mol.

As a general conclusion, it follows that the heat of formation of Fe_3C lies within the range +8 and −4 K cal./mol. Its exact value depends on the condition of the carbon, and decreases with increasing graphitisation.

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[3] H. Esser and W. Bungardt, "Heat Change of the Austenite-martensite Transformation," March, 533–536.
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[5] F. Wever and G. Naeser, "Calorimetric Investigations of the Tempering Reactions in Hardened Carbon Steels," Mitt. a.d. K.W. Inst. f. Eisenforschung, 1933, 15, No. 4, 37–47.
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Books and Publications.

The Alloys of Iron and Tungsten. By J. L. GREGG. Pp. xii and 511. Published for the Engineering Foundation by McGraw-Hill Book Company, Inc. Price 36s. net.

THE Iron Alloys Committee of the Engineering Foundation was instituted in the United States in 1929, with the laudable object of making a survey of the existing literature relating to iron and its alloys, so as to provide producers and users of these materials, as well as research workers and others who may be interested, with summaries of basic data and comprehensive bibliographies. It is not an object of the Committee to initiate and support researches for the discovery of new products in this field of metallurgical activity, although investigations may be organised, if deemed desirable, to fill in gaps which the survey of the literature might reveal. The work is under the direction of Dr. F. T. Sisco, and already volumes have appeared dealing with alloys of iron and molybdenum under the same authorship as the present monograph, and with alloys of iron and silicon by E. S. Greiner, J. S. Marsh, and B. Stoughton.

No one will dispute the great value of bibliographical research, of which this volume is such an excellent example. The technical literature of the world is vast, and it is most necessary that its several branches should, from time to time, be summarised. To a limited extent this is frequently done in published researches, but usually only for some specialised part of the subject. It is to be hoped that the publication of these most useful monographs by the Engineering Foundation will encourage the production of other works, for there can be no doubt that in summarising the stage of development of any section of metallurgical science and industry and providing a carefully selected bibliography, a very great service is rendered to the industry and to those responsible for its further development.

It would materially assist in achieving the object for which books of this kind are expressly prepared if they were made as short as possible, without the sacrifice of any vital matter, to permit of their production relatively cheaply and so ensure the widest possible circulation. In this matter of cost it is not to be forgotten that such works as this rapidly become out of date. The text of Dr. Gregg's book, which includes no fewer than 184 figures and eighty tables, consists of sixteen chapters covering 446 pages. For a highly specialised subject, many of the problems of which are still in a very unsettled and controversial state, this is indeed a full summary. It could with advantage have been shorter, for much of the detail which is given is of a kind in which perhaps only the research worker, who in any case would consult the original references, is interested. The author, however, is to be congratulated on having produced a very thorough and well-balanced survey of the subject. No fewer than 1800 articles and papers were consulted, and of these more than 1300 were abstracted. After such a searching examination of the literature it is possible to eliminate duplicate and secondary articles, and appended to the text is a selected bibliography of 566 papers covering the years 1860 to 1933.

The first chapter is of a general character and is concerned with the development and use of tungsten alloys, the occurrence of tungsten ores and the production of metallic tungsten and ferro-tungsten. The next four chapters deal with iron-tungsten, iron-carbon, and tungsten-carbon alloys, and the ternary alloys of iron-tungsten-carbon, the constitution of which, as derived from thermal, microscopic, magnetic and X-ray studies, is discussed at length. The remaining chapters are of a less theoretical nature and are concerned with the production and properties of ferrous alloys containing tungsten. The manufacture of tungsten steels is covered in a short chapter, which is followed by one dealing with the characteristics of these alloys, in which are discussed the effect of tungsten on transformation temperatures and on the heat treating range, and the physical and chemical properties of tungsten steels as well as the miscellaneous effects of tungsten additions. Succeeding chapters deal with the mechanical properties of tungsten steels, tungsten magnet steels, and the effect of tungsten in cast steel and cast iron. Steel users will be especially interested in the very informative chapter devoted to tungsten in tool and die steels and the three chapters on the nature of high-speed steels, and the cutting and other properties of these alloys. This most useful and comprehensive study and summary, which anyone interested in ferrous alloys containing tungsten will find of great value, is concluded with a chapter concerning corrosion and heat-resisting alloys and alloys of the Stellite type.